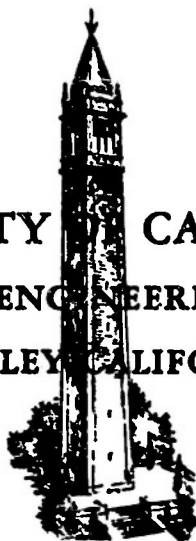


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EFFECT OF DISPERSIONS ON MECHANICAL PROPERTIES

Thirty Third Technical Report

By

John E. Dorn and C. Dean Starr

22, N7-onr-295, Task Order II,
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BEST SCAN AVAILABLE

November 9, 1953

Office of Naval Research
Department of the Navy
Washington 25, D. C.

ATTENTION: Mr. Julius Harwood

Dear Sir:

Attached hereto is a copy of the Thirty Third Technical Report by John E. Dorn and C. Dean Starr. The American Society of Metals requested that we prepare a paper to be included in a seminar on the effect of microstructure on the mechanical properties of metals and alloys. This report entitled "Effect of Dispersions on Mechanical Properties" contains a review of existing data previously obtained either in our laboratories or from the literature.

Respectfully submitted,



John E. Dorn
Professor of Physical Metallurgy

JED:dk

EFFECT OF DISPERSIONS ON MECHANICAL PROPERTIES

By

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C. Dean Starr⁽²⁾

Thirty Third Technical Report, Series 22, Issue 33
N7-onr-295, Task Order II, NR-031-048

November 15, 1953

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INTRODUCTION

Scientific interest in the effect of microstructure on the mechanical properties of metals had its origin with the development of the first metallurgical microscope. As early as 1722 Reaumur⁽¹⁾ recorded the results of his investigations on the metallography and mechanical properties of iron in his monograph entitled "The Transformation of Wrought Iron into steel". Although Reaumur did not attempt any direct correlations between the microstructure and the strength and ductility of irons, he nevertheless suspected that such correlations might exist. It was not until about 1887 that the first definitive example of the dependence of the mechanical properties of metals on their metallographic structure was recorded. In that year Sorby⁽²⁾ presented a written discussion on the microscopical structure of iron and steel in which he ascribed the greater softness of annealed grey iron castings as contrasted to hardness of chilled castings to the replacement of cementite by graphite flakes. Additional investigations by Sorby on cast steel, malleable cast iron and meteoric iron served to confirm the thesis that the properties of metals could be correlated qualitatively with their microstructure. Over the intervening years extensive investigations have clearly established the fact that the properties of polyphase alloy aggregates depend not only on their composition but also on the kind, shape, percentage and distribution of the various phases present in the alloy.

A brief summary of our knowledge of the dependence of properties on the microstructures of polyphase alloys will be attempted in this review. At present the correlations that can be achieved are substantially qualitative; it is not yet possible to predict the properties of polyphase alloys

from a knowledge of the properties of the individual phases and the micro-structure. Several areas wherein additional investigations are required to provide the missing concepts for accurate predictions will be described. But no attempts will be made here to discuss the complex problems of precipitation hardening which is outside the realm of this paper inasmuch as it involves strengthening of the matrix itself due to coherency stresses.

POLYCRYSTALLINE METALS

An understanding of the plastic behavior of polycrystalline aggregates is essential to the complete understanding of the plastic behavior of polyphase alloys. This arises because limited volumes in the polyphase alloys consist of a polycrystalline aggregate of a single phase. Furthermore the interactions between two adjacent grains of two dissimilar phases in a polyphase alloy is in effect a more pronounced example of a similar interaction between two adjacent grains of the same phase one of which is favorably and the other of which is unfavorably oriented for deformation under the applied stress.

Carpenter and Klam⁽³⁾, Aston⁽⁴⁾, Yamaguchi⁽⁵⁾ and many others have demonstrated that coarse grain polycrystals usually exhibit less deformation in the vicinity of the grain boundary than in the core of the grains. A short time ago Boas and Hargreaves⁽⁶⁾ appeared to have uncovered some contradictory evidence. They made a series of Vicker's diamond hardness impressions along a line parallel to the tensile axis on a polished surface of a coarse grained polycrystalline aggregate of a commercial aluminum specimen. After tensile straining it was found that the local strains between the hardness impressions varied not only from grain to grain but also within a single grain. In general the strains that were measured across the grain boundaries were about the average of those exhibited by

the cores of the two adjacent grains. This suggested the weaker more favorably oriented grain induced the adjacent stronger and less favorably oriented grain to deform in the boundary region and conversely the stronger grain inhibited the deformation of the weaker grain in the boundary region. More recently, however, Urie and Wain⁽⁷⁾ have repeated these investigations using shorter microgage lengths obtained by photogripping high purity aluminum specimens. They found that the stronger as well as the weaker grains exhibited less deformation very near to the grain boundaries thereby reconfirming the previous observations that in general grain deformations are restrained in the region of a grain boundary.

During the years when popular opinion suggested that the grain boundaries consisted of an amorphous cement, the greater resistance of the grain boundary region to deformation was partly ascribed to the higher low-temperature strength of the amorphous cement in the boundary. In the following years of decline of the amorphous cement hypothesis more emphasis was placed on the thought that the resistance to deformation in the boundary regions of two dissimilarly oriented grains arose from their mutual resistance to slip. The proof of this concept was presented by Chalmers⁽⁸⁾ in a series of unique tests. Chalmers produced a series of cylindrical bicrystals of Sn having a longitudinal boundary. The bicrystals were so grown that the $[001]$ axes of both crystals were normal to the specimen axis whereas their $[101]$ axes were about 45° to the specimen axis. In this way a series of bicrystals were produced in which the angle between the $[001]$ axes of the two crystals could be varied from 0° to 90° . Thus each half of each bicrystal experienced the same resolved shear stress for slip at the same tensile stress.

The results of this investigation, as shown in Fig. 1, reveal that the stress required to produce a definite strain varied linearly with the

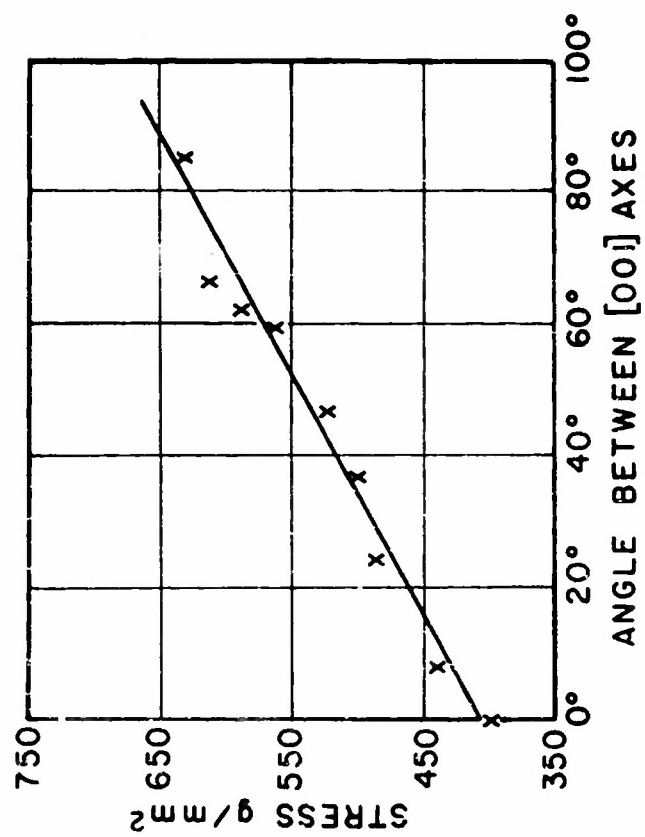


FIG.1 EFFECT OF ORIENTATION ON THE YIELD STRESS
OF TIN BICRYSTALS (AFTER CHALMERS)

angle between the $[001]$ axes, being a minimum for 0° and a maximum for 90° . Since an extrapolation of the deformation stress of the bicrystals to 0° gives the deformation stress of a single crystal, Chalmers' data suggest that the mechanical properties of these bicrystals are a function of the difference in orientation between two conjugate crystals and are independent of the grain boundary per se. Otherwise, a simple linear extrapolation of the data for bicrystals to 0° between the $[001]$ axes would have given a higher deformation stress than that for the single crystal. Thus it appears that the mechanical properties of a polycrystalline aggregate are also a function of the relative orientations between contiguous grains.

Unfortunately Chalmers' experiments are not sufficiently general to reveal all of the factors responsible for the plastic properties of polycrystalline aggregates. If the resistance to deformation of a group of grains were exclusively determined by their mutual orientations, a larger specimen containing a geometrically similar set of grains should exhibit the same deformation stress as a smaller specimen containing a set of proportionately smaller grains. But as shown by the typical example recorded in Fig. 2A, the deformation stress depends on the mean grain diameter of the polycrystalline aggregate⁽⁹⁾.

The results of an interesting investigation by Pell-Walpole⁽¹⁰⁾ on the influence of specimen size on the effect of grain size on the ultimate tensile strength of "pure" tin are recorded in Fig. 2B. Specimens A and B were prepared to the same width (0.5 in.) and differed only relative to their thicknesses which were 0.1 in. and 0.2 in. respectively. The specimen having the greater cross-sectional area exhibited the greater ultimate tensile strength over the range of large grain sizes illustrating that a definite size effect exists, when the grains over the cross-section area are fewer than about 20. Over this range a large effect of increasing

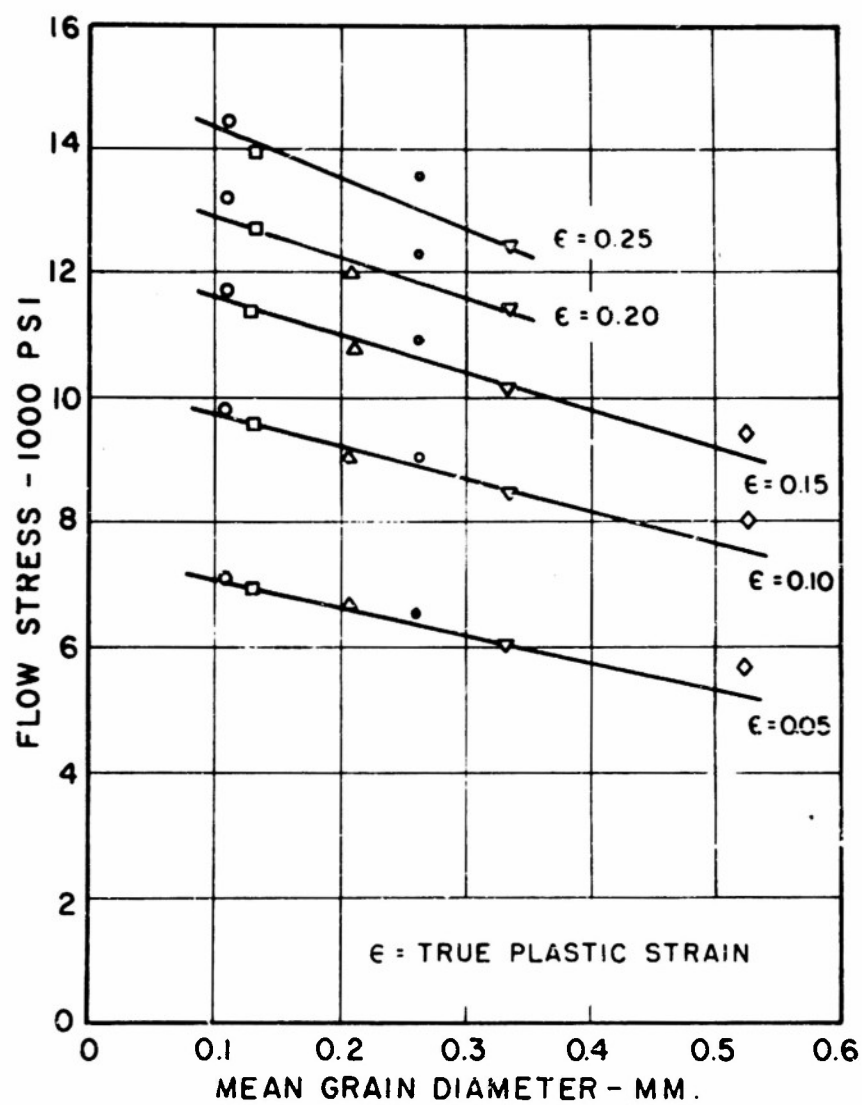


FIG. 2A EFFECT OF GRAIN SIZE ON FLOW STRESS OF
PURE ALUMINUM AT 194° K.
(AFTER DORN, PIETROKOWSKY, AND TIETZ)

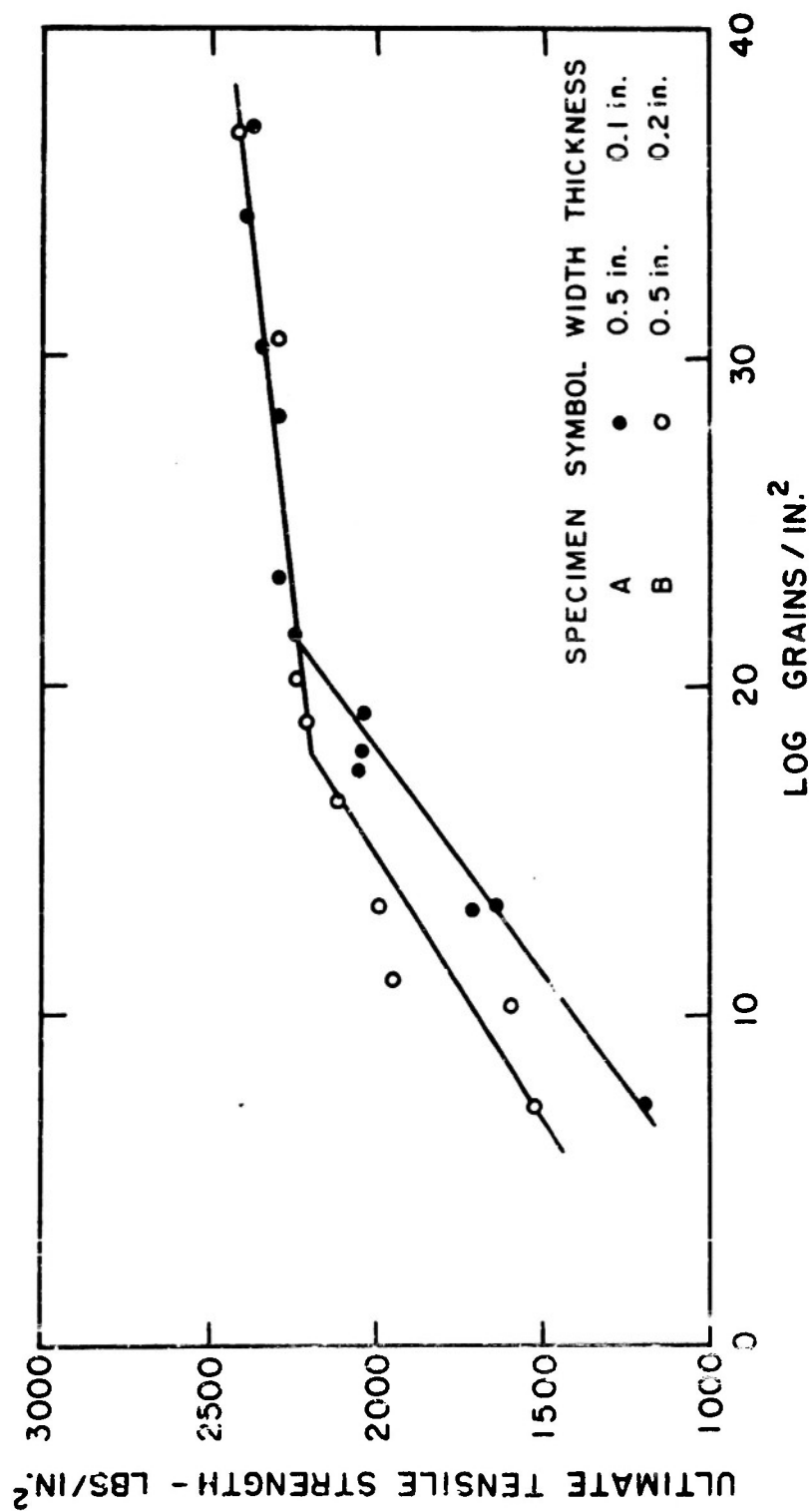


FIG. 2B RELATION BETWEEN GRAIN SIZE AND TENSILE STRENGTH OF PURE TIN FOR TWO CROSS-SECTIONS. (AFTER PELL-WALPOLE)

tensile strength with decreasing grain size was observed. But when the number of grains over the cross-section exceeded about 20, no specimen size effect was obtained and the ultimate tensile strength increased more slowly with a decrease in grain size. The persistence of the small but consistent increase in ultimate tensile strength with decreasing grain size, beyond the point where the specimen cross-sectional area is also pertinent, suggests that not only relative orientation but also grain size affects the plastic properties of metals.

Two separate factors probably contribute to the commonly observed grain size effects. First recrystallized specimens of some metals exhibit lower strengths following annealing at higher temperatures even when no change in grain size is detected. This fact suggests that structural changes leading to softening might occur within the recrystallized grains during annealing. Consequently part of the softening obtained after annealing in the range where grain coarsening is obtained, might be ascribed to this phenomenon whereas the residual softening might be ascribed to grain coarsening.

The failure of the laws of geometrical similarity, when applied to grain size, suggests that some of the geometrical features of interference to migration of dislocations in the vicinity of the grain boundaries are not geometrically similar in specimens of different grain sizes.

This thesis is supported by the recent investigations of Jaswon and Foreman⁽¹¹⁾ who have deduced that the equilibrium distance of a dislocation from a grain boundary is independent of grain size.

Perhaps this factor contributes to the failure of the laws of geometrical similarity when applied to grain size effects on the plastic properties of polycrystalline aggregates.

In some respects the plastic behavior of polyphase alloys cannot be

too dissimilar from that of a polycrystalline aggregate of a single phase system. Several investigators have shown that, in an alloy containing two ductile phases, the softer phase deforms first and subsequently both phases exhibit slip. This is analogous to the common observations made on polished specimens of polycrystalline aggregates of a single phase alloy. For example during deformation of high purity magnesium, certain favorably oriented grains will show slip bands or mechanical twinning first. Other less favorably oriented grains will exhibit slip bands or twinning at some higher stress level following more extensive straining and consequent strain hardening of the more favorably oriented grains. Such similarity in behavior might suggest that Taylor's⁽¹²⁾ synthesis of the plastic properties of polycrystalline aluminum from the known plastic behavior of single aluminum crystals might be extended to alloys containing aggregates of several ductile phases.

Taylor's analysis was based on the following assumptions:

1. Under homogeneous deformations, each grain of the polycrystalline aggregate undergoes the same strain as the macroscopic strain.
2. Each grain exhibits the same slip mechanisms as a single metal crystal.
3. Strain hardening is the same under multiple slip as under single slip.
4. Since the six components of the strain tensor are related through the single equation (Equation of Continuity) for constancy of volume, five mechanisms of slip must be invoked for general deformation.
5. The five active slip mechanisms will be those for which the plastic strain energy is a minimum.

On the basis of these assumptions Taylor calculated the stress-strain curve for a randomly oriented polycrystalline aggregate of aluminum and

predicted the orientation changes that would be experienced by each differently oriented grain. The excellent correlation that was achieved between the calculated and experimental stress-strain curves for polycrystalline aluminum seemed at first to confirm this method of analysis. But subsequent x-ray investigations by Barrett and Levenson⁽¹³⁾ revealed that about one third of the grains reorient in ways contradictory to Taylor's predictions. Furthermore the remarkable agreement that was obtained between the predicted and experimentally determined stress-strain curve probably arose from accidental and fortuitous selection of a specimen with an appropriate grain size to achieve correlation.

Undoubtedly the deviations between the theoretical predictions and the experimental facts arose from the simplifying assumptions that were made in order to reduce the monumental task of analysis:

1. Numerous microscopic investigations have shown that each grain of a polycrystalline aggregate does not experience the same strain.
2. Furthermore Barrett and Levenson have shown that the strains are highly heterogeneous even in a single grain as evidenced by deformation banding.
3. The heterogeneous nature of the deformation processes suggests that the stresses also vary locally over a single grain. In fact recent observations have shown that in one region of a grain in a polycrystalline magnesium specimen may exhibit slip bands whereas in another region of the same grain no slip bands can be detected.⁽¹⁴⁾
4. Furthermore grain size plays an important role in the plastic behavior of polycrystalline aggregates.

In view of the failure of the laws of geometrical similarity when applied to grain size effects, it must be presumed that the pattern of trapped dislocations in the grain boundary region deviates from similarity.

Consequently deformation banding might be sensitive to grain size as is the deformation strength and the rate of strain hardening. These are some of the factors which must be more fully explored before a completely rational approach to the properties of polyphase alloys can be formulated.

TWO DUCTILE PHASES

At present only the crudest types of analyses are available for estimating the plastic behavior of an aggregate of two ductile phases from the plastic properties of each individual phase. In an attempt to preserve continuity one might assume Taylor's postulate that the strains are identical over all grains of each phase. If this assumption be applied the stresses must be dissimilar in each of the two different phases. If for example the flow stress is σ' for the first phase and σ'' for the second phase at the same strain ϵ' , the average stress over the alloy will be given by

$$\sigma = f'\sigma' + f''\sigma'' \quad (1)$$

where f' and f'' are the volume fractions of the two phases. In this way the stress-strain curve for 50 volume percent of each phase, as shown by the broken line of Fig. 3, might be calculated. In this case, the stress for a given strain would increase linearly with the volume fraction of the stronger phase.

But, as previously mentioned, the assumption that each phase undergoes the same strain deviates from the facts and becomes seriously in error as the differences in the flow stresses of the two phases increase. Perhaps in an attempt to preserve macroscopic equilibrium of stresses, the alternate assumption that each phase is subjected to the same stress might be made. If ϵ' and ϵ'' are the strains experienced by the each of

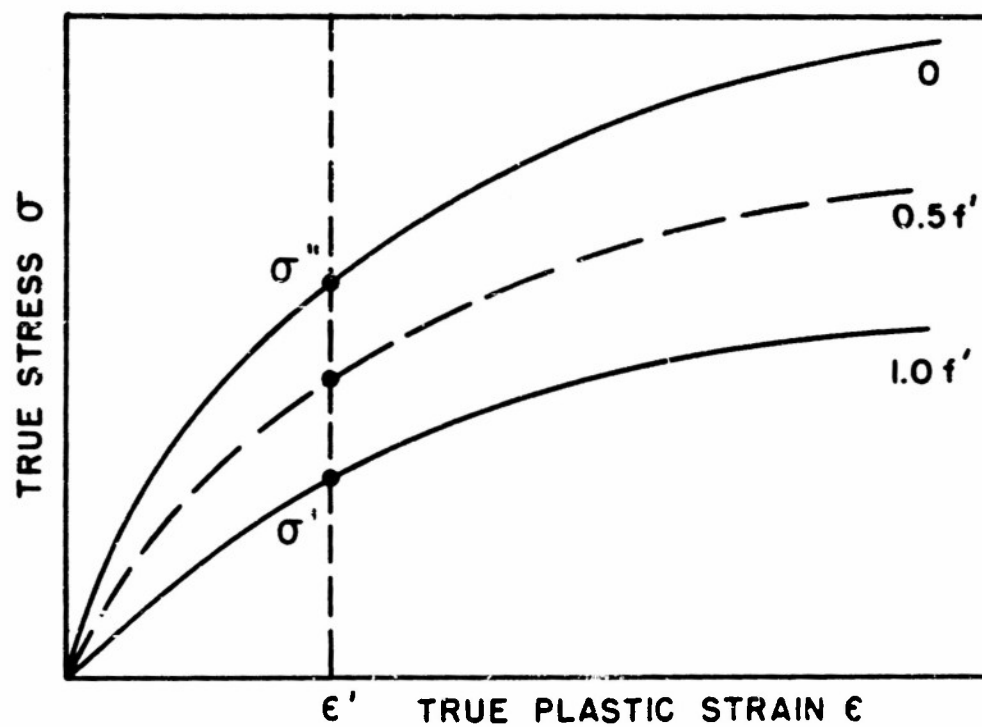


FIG. 3 ESTIMATION OF THE STRESS-STRAIN CURVE
BASED ON THE ISO STRAIN HYPOTHESIS.

the two individual phases at stress $\bar{\sigma}$, the average strain of the aggregate would be given by

$$\bar{\epsilon} = f' \epsilon' + f'' \epsilon'' \quad (2)$$

where the strain curve for 50 volume percent of each phase would be calculated as shown in Fig. 4. In this example the flow stress for a given strain would no longer be a linear function of the volume fraction of the phases. However the simple assumption that each grain of each phase is subjected to the same stress is also untenable, because this assumption neglects the requirements for continuity. A crude guess might suggest that a better estimate of the stress-strain curve might be approximated by the mean values obtained from the isostrain and the isostress hypothesis.

Although numerous investigations have been made on the plastic properties of simple alloy systems, neither the property measured nor the data reported permit a numerical check on the above approximations^(15,16). Evidence is available which suggests that the deformation of ductile polyphase alloys is partially isostrain and partially isostress. This is so because the two conditions of equilibrium of stresses and continuity of strains across the two phase boundary must be satisfied simultaneously.

An interesting example of plastic flow in an alloy containing two ductile phases was observed in the investigation conducted by Boas and Honeycombe⁽¹⁷⁾ on the deformations in α - β brass at room temperature. Microscopic examination of a polished surface of the specimens revealed that slip lines were evident in the α phase after 1% total extension while the β brass exhibited no detectable deformation. After 10% deformation, extensive slip and surface rumpling had occurred in the α phase while only slight visible slip was evident in the β phase. Accordingly, the deformation of two phase alloys at least partially coincides with the isostress assumption.

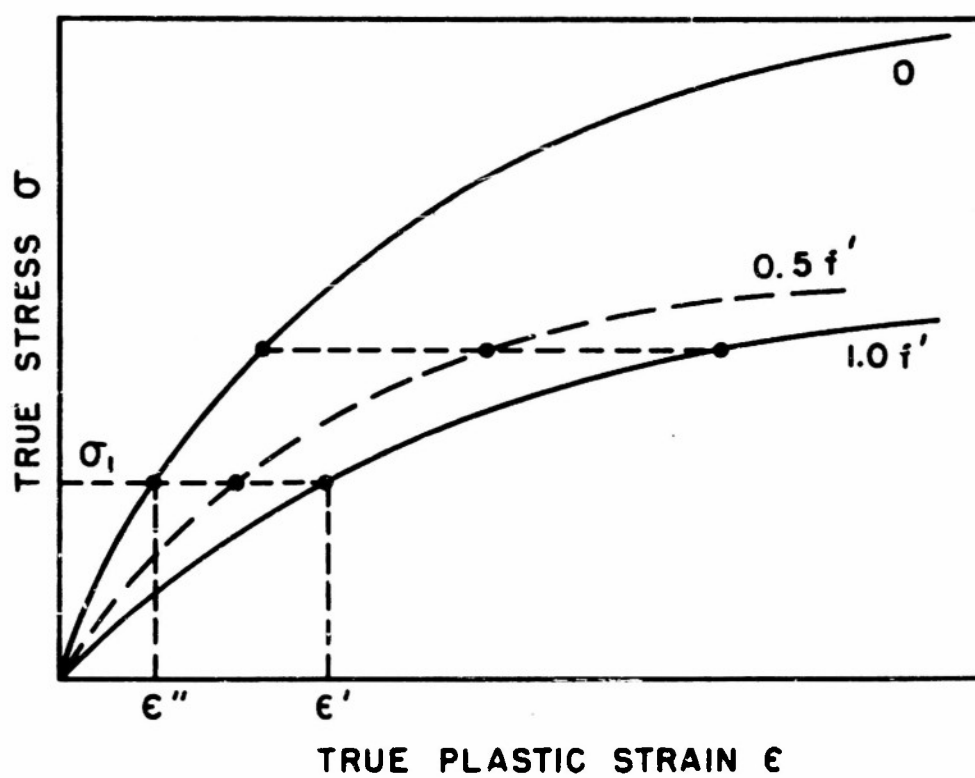


FIG. 4 ESTIMATION OF THE STRESS-STRAIN CURVE
BASED ON THE ISOSTRESS HYPOTHESIS.

Since deformation initially occurs only in the softer of the two phases present, high stresses are set up in the region near the boundary between the phases. If the stress concentration is insufficient to exceed the yield point of the harder phase, additional deformation will occur in the softer phase. But with ever increasing applied stresses, the stress level will approach that of the yield point of the harder phase. Under such conditions, the softer phase might induce the harder to deform near its boundary while the harder phase restricts the deformation of the softer phase in the vicinity of the boundary. This effect is in partial agreement with the isostrain hypothesis. These views have been confirmed not only by the microscopical observations on brass by Boas and Honeycombe but also on Ag-Mg alloys by Clarebough⁽¹⁸⁾. He observed that the harder β phase was more deformed near the boundary than in the center of the β crystals, and the deformation of the softer α phase was smaller in the vicinity of the β crystals. Moreover the harder β phase deformed more at the same stress in the presence of increasing amounts of the softer α phase.

Extensive investigations have been conducted on the hardness of a series of binary alloys. A typical example of the results of such investigations is reproduced from the data of Meyer⁽¹⁹⁾ given in Fig. 5. As cast Pb-Sn alloys exhibit an increase in hardness with increasing percentages of the Sn phase over the two phase region. The apparently abnormal peak in hardness obtained in the vicinity of the eutectic composition has been ascribed to the finer grain sizes introduced by the eutectic reaction. As expected this peak is less pronounced when casting is done in preheated molds. Annealing at 178°C and subsequent quenching results in solution strengthening of the Pb rich and Sn rich phases so that both phases exhibit about the same hardness. Aging at room temperature gives a hardness curve that is linear with the volume fraction of the two phases in fair agreement

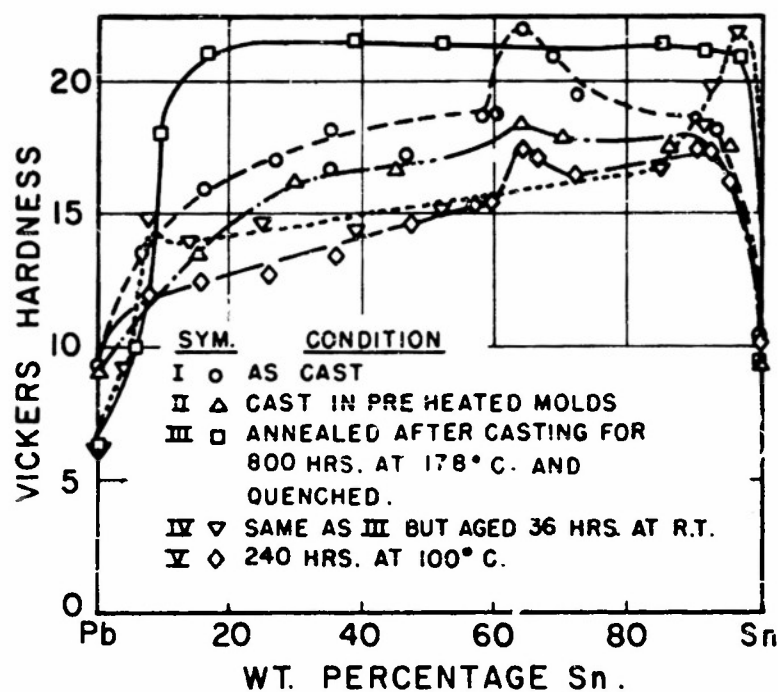
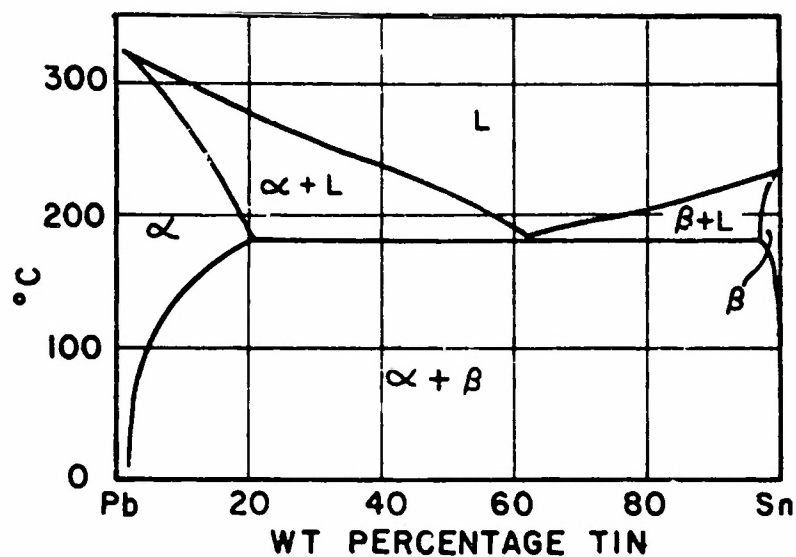


FIG. 5 EFFECT OF COMPOSITION ON THE HARDNESS OF Pb-Sn ALLOYS.

with the predictions based on the previously described isostrain hypothesis. The abnormal hardness peak in the Sn rich region of the diagram has been ascribed to the production of a fine precipitate of the Pb rich phase in the Sn rich phase. Aging for 240 hrs. at 100°C results in agglomeration of the fine precipitates with consequent removal of the hardness peak in the Sn rich region. Except for retention of the hardness peak near the eutectic composition the hardness increases linearly with the volume fraction of the Sn rich phase over the two phase region.

HARD-DUCTILE PHASE

In general, hard-ductile phase alloys can be classified into three types, namely:

1. Continuous brittle phase surrounding the ductile phase.
2. Discontinuous brittle phase in the grain boundaries of the ductile phase.
3. Discontinuous brittle phase distributed randomly in the ductile matrix.

Typical examples of the first classification are Au-Bi and oxygen free Cu-Bi alloys. Arnold⁽²⁰⁾ in his investigation on Au-Bi alloys observed that the alloys were very friable, but the gold crystallites which separated were about as ductile as native gold. He concluded that the bismuth formed a continuous brittle film around the gold crystallites, which causes the alloy to be brittle even though the Au was extremely ductile. Similar observations have been made on Cu-Bi alloys and extensive investigations have led to the conclusion that the brittleness of these alloys is attributable to either a continuous boundary region rich in bismuth, or to a continuous submicroscopic film of bismuth⁽²¹⁾.

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In the presence of oxygen the bismuth in copper agglomerates and precipitates in the grain boundary as globules of bismuth oxide which produces a discontinuous boundary precipitate and greatly reduces the brittleness of Cu-Bi alloys⁽²²⁾. Discontinuous brittle phases in the grain boundary have also been observed in the Cu-Sb system⁽²³⁾, in internally oxidized Ag, Cu, and Ni containing solutes having a great affinity for oxygen,⁽²⁴⁾ and in zinc die castings containing impurities such as Cd, Pb, Sn⁽²⁵⁾. Microscopic examination of Cu-Sb alloys by McLean and Northcott showed discontinuous precipitates in these alloys. Oxidized samples of Ag containing small amounts of Mo and Al, Cu containing Mn, and Ni containing Al exhibited an increase in tensile strength but a decrease in elongation which suggested the occurrence of oxides in the grain boundaries. Critical experiments on oxidized single crystals of the same composition showed that the crystals were quite ductile. Accordingly, it was concluded the oxides were present as discontinuous precipitates in the grain boundary.

It is well known that ductile zinc die castings become extremely friable and fracture readily when Cd, Pb or Sn are present. The assumption that discontinuous precipitates are present in the grain boundaries appears substantiated by the indirect evidence that Zn alloys containing high Cd, Pb or Sn are susceptible to intergranular corrosion.

DISPERSIONS

Most two or more phase alloys are of type three wherein the alloy contains a hard phase or particle distributed randomly in a continuous ductile phase. Typical examples are particles of Fe_3C in a matrix of alpha iron, and CuAl_2 particles in a solid solution of copper in aluminum.

Microscopic examination of these alloys indicate that plastic deformation occurs only in the ductile matrix.

The significant effect of dispersions of hard constituents embedded in a soft ductile phase on the plastic properties of polyphase alloys was uncovered shortly following the introduction of metallurgical microscopy, early in the history of scientific physical metallurgy. Over the intervening years, extensive qualitative investigations have been made which have permitted the improvement of the yield strengths of commercial alloys merely by refining the dispersion of the hard constituent in the soft ductile phase. But in spite of the recognized technological importance of this subject, no systematic study of the effects of dispersions on the plastic properties of alloys was made until 1942 when Gensamer and his colleagues first reported their investigations on the effect of dispersions of carbides in ferrite on the plastic properties of steels^(26,27). A summary of their essential observations on the effect of dispersions on the deformation strengths of a series of steels is reproduced in Fig. 6 where the deformation strength at a plastic strain of 0.20 is given as a function of the average distance between carbide particles dispersed in the ferrite phase. The recorded data apply to various steels that had about the same percentage of minor chemical components and differed principally relative to their carbon content and heat treatment. Whereas the pearlite was produced by isothermal decomposition of austenite, the spheroidites were obtained by tempering martensite. These data reveal that, within a rather broad scatter band, the flow strength of dispersions of carbides in ferritic steels decreases linearly with the logarithm of the mean ferrite path independent of carbon content or type of dispersion, be

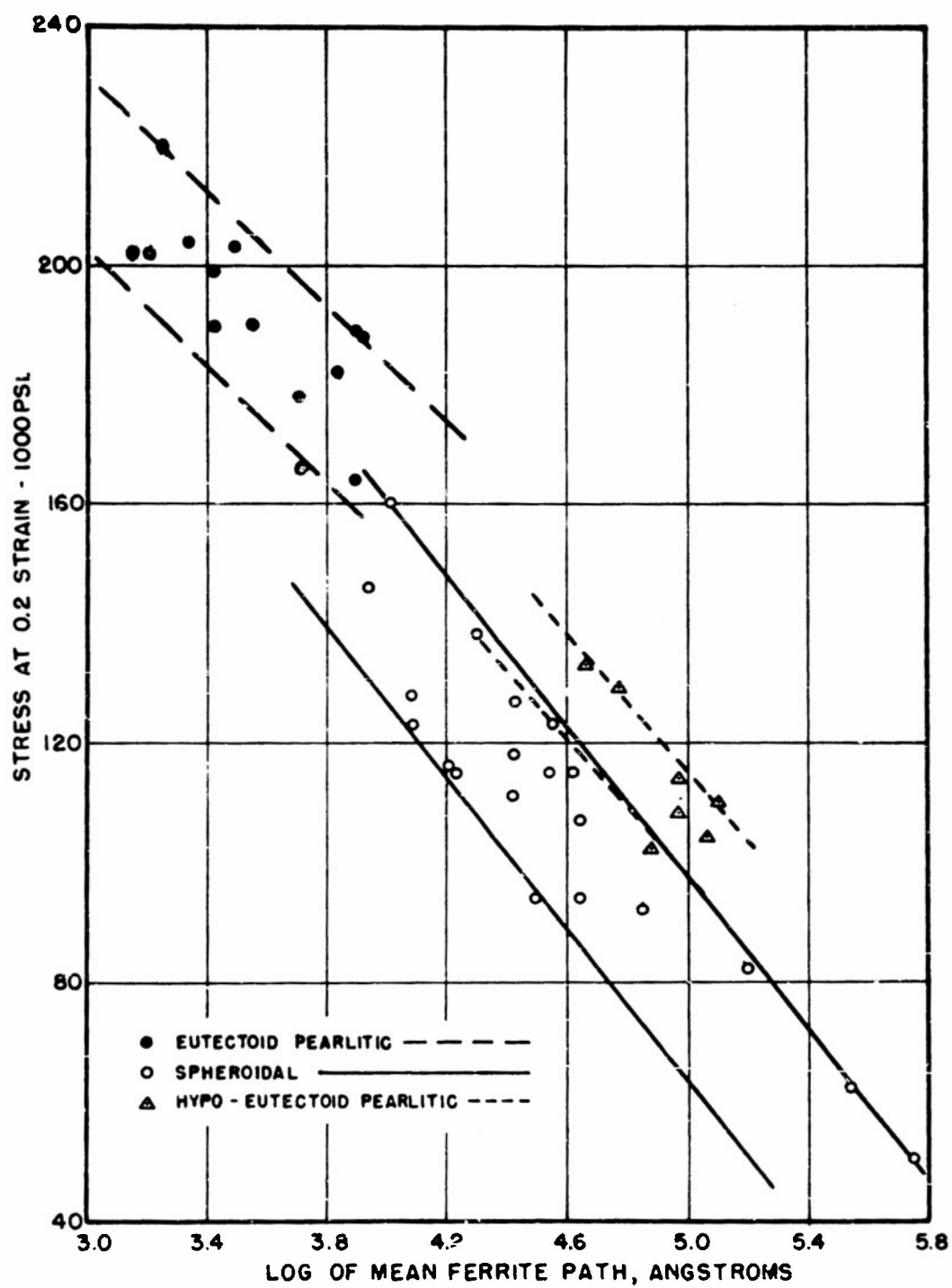


FIG. 6 THE EFFECT OF THE MEAN DISTANCE BETWEEN PARTICLES
ON THE FLOW STRESS OF STEELS.
(AFTER GENSAMER)

it entirely lamellar, lamellar islands in ferrite, or spheroidal. When the individual scatter bands for eutectoid pearlitic, hypoeutectoid pearlitic, and the various spheroidal dispersions are separately demarked, as shown in Fig. 6, however, the spheroidal dispersions appear to give slightly lower deformation strengths at strains of 0.20 than is given at the same mean free ferrite path for the pearlitic types of dispersions. In view of the recognized difficulties in measuring mean ferrite paths this difference might be ascribed to scatter in the data.

More recently Roberts et al.⁽²⁸⁾ extended the investigations originated by Gensamer and reported the results given in Fig. 7. Their measurement of the mean free path in the hypoeutectoid pearlites and puron, however differed from that used by Gensamer insofar as they measured the mean spacing between the pearlite patches, ferrite grain boundaries, and inclusions. This choice was probably based on the hypothesis that the ferrite in the pearlite patch does not deform under the small total plastic strains at the lower yield point. This assumption is probably incorrect for eutectoid pearlites.

Although the semi-logarithmic relationship must fail for large mean free paths, it is again shown to be valid for paths down to about 5000°A and the data appear to indicate that the shape of the hard particles has at most a minor effect on the yield strength of dispersion alloys. Both Gensamer's and Roberts' data suggest that the deformation strength of steels decreases linearly with the logarithm of the mean ferrite path independent of the type of dispersion and carbon content. But as will be presented later, real differences in the plastic properties of pearlitic and spheroidal dispersions appear to exist.

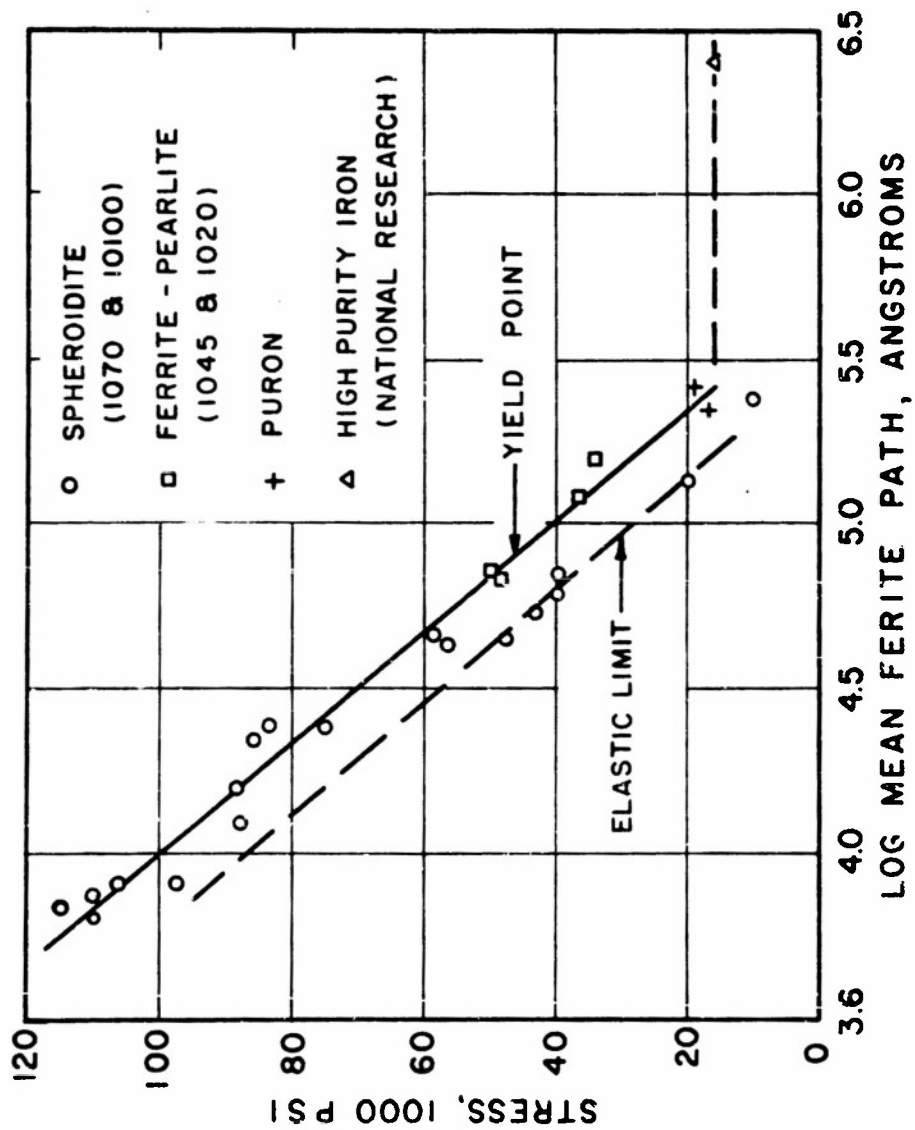


FIG. 7 EFFECT OF DISPERSIONS OF Fe_3C ON THE YIELD STRENGTH OF STEELS.

It is well known that a wide variety of steels obey the empirical stress-plastic strain relationship

$$\sigma = A \epsilon^m \quad (3)$$

where σ = true deformation stress
 ϵ = true plastic strain
 A = the deformation stress at unit strain
 m = the strain hardening index.

Since A and m are constants for any given steel, they serve to characterize its plastic properties. If, for a given class of steels, m is a single-valued function of A , that class of materials is said to provide an homologous series of stress-strain curves. Although A might increase from specimen to specimen in that class, m will also change systematically so that an entire homologous family of stress-strain curves are obtained in the homologous series of steels. Conversely, if m is not a single-valued function of A within a group of steels, the group will not be homologous. Obviously the plastic properties of dispersions can be correlated uniquely with the mean ferrite path only within an homologous series.

Reference to Fig. 8 reveals that, in general, the strain hardening index decreases uniformly with increasing values of the deformation strength at a strain of 0.20 for a wide variety of steels. This suggests that within the scatter band obtained these steels form an homologous series of alloys. But upon closer inspection of the data, certain well-documented differences between the various types of steels become evident. For example, the strain hardening index, m , for bainites is greater than that for manganese steel pearlites at the same deformation strength. Although the initial analysis suggested that all of the steels represented form an

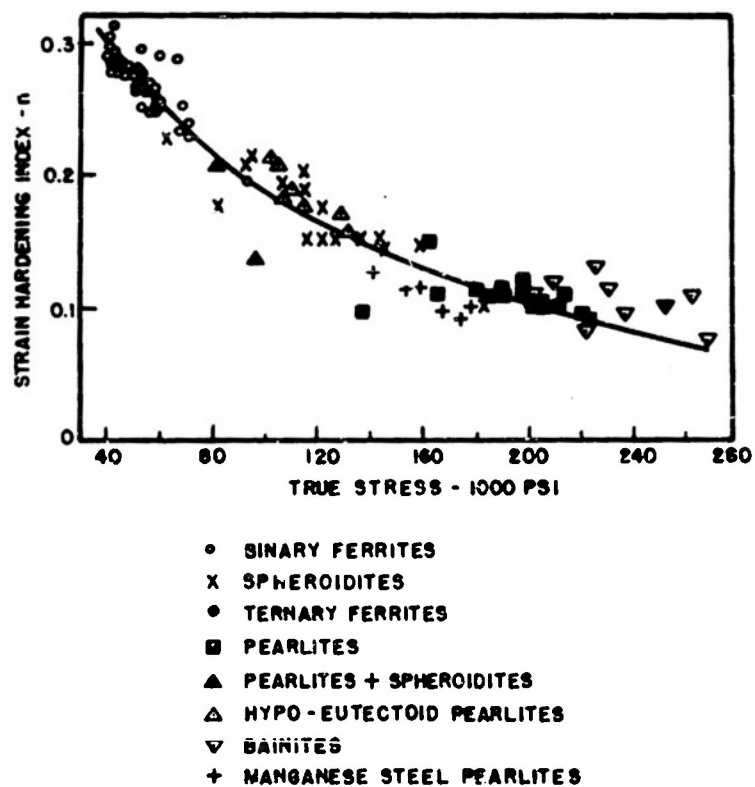


FIG. 8 CORRELATION OF STRAIN HARDENING INDEX, n ,
WITH THE DEFORMATION STRESS AT A STRAIN OF 0.20.
(AFTER GENSAMER)

homologous series of alloys, a more refined analysis suggests that several groups of steels represented in Fig. 8 are somewhat unique.

Bain's data⁽²⁹⁾ on the effect of dispersions on the plastic properties of an eutectoid steel are given in Fig. 9. These data clearly reveal that, for the same hardness values, the proof stress, breaking stress and tensile strength for the pearlitic structures are distinctly different from those for the spheroidal dispersions. Consequently, the pearlitic and spheroidal dispersions do not give a single homologous series of stress-strain curves. These deductions insist that the type of structure as well as its dispersion must influence the plastic properties of polyphase alloys. Thus the plastic properties of an alloy containing a hard phase dispersed in a soft ductile phase cannot depend only on the mean path between the hard constituents in the ductile phase but it must also be dependent on the type of structure. The investigations by Gensamer et al. and Roberts et al. reveal that the mean free ferrite path is the more important factor for correlating the plastic properties of steels, and that it is probably the unique factor for a given type of structure.

The plastic properties of nonferrous alloys have also been investigated^(30,31). In these investigations, the influence of CuAl_2 dispersions in a solid solution matrix of Cu dissolved in Al on the plastic properties were determined over a temperature range of 78°K to 700°K. The chemical composition, grain size, and mean free path between the CuAl_2 particles in the alloys investigated are recorded in Table I. Special heat treatment were employed to obtain fairly random spheroidal dispersions dispersed in an alpha solid solution of aluminum of constant grain size. Since the details of the heat treatment, testing procedure and method of estimating the volumetric and planar mean free paths are recorded elsewhere, they

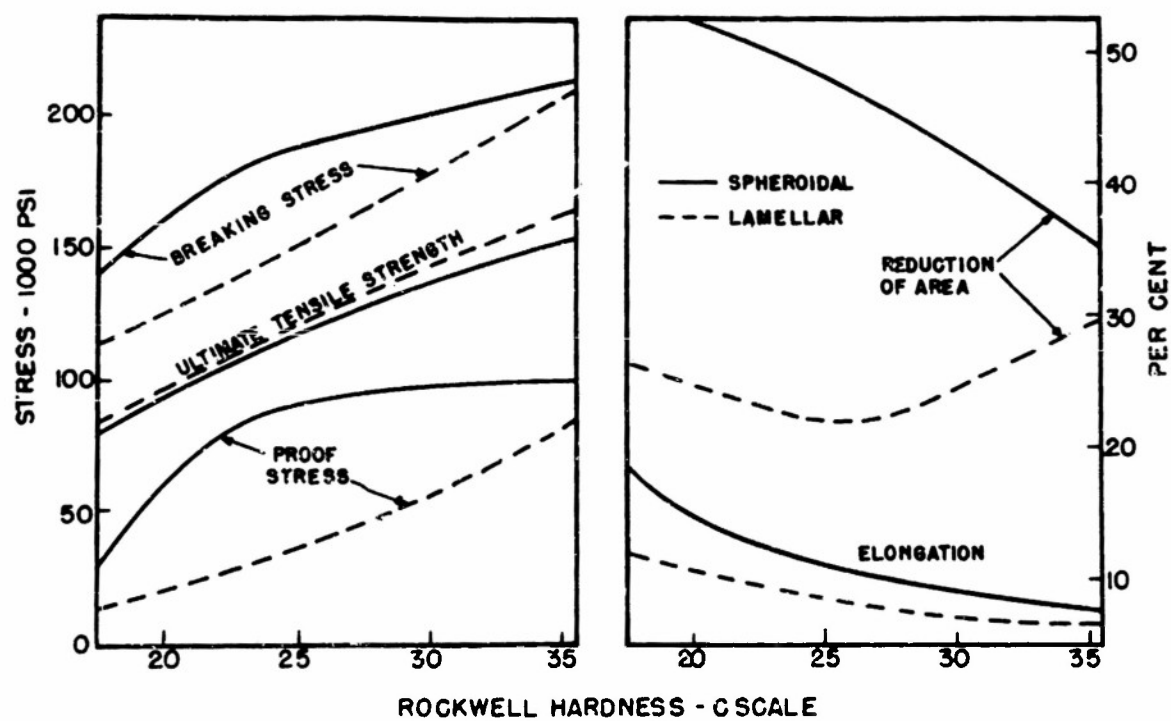


FIG. 9 THE INFLUENCE OF STRUCTURE ON THE TENSILE PROPERTIES OF A EUTECTOID STEEL .
(AFTER BAIN)

TABLE IChemical Composition, Grain Size and Volumetric Mean Free Path

Dispersion	Grain Size in grains/mm	V.M.F.P.	Composition by Weight*				
			Chemical Analysis	Spectro-Chemical Analyses			
				Si	Fe	Mn	Zn
3% M*	2.4	0.00096	3.05	0.002	0.002	trace	0.0004
3% C	2.3	0.00365					
4% F	2.3	0.00078	4.03	0.002	0.002	0.001	0.0004
4% M	2.3	0.00155					
4% C	2.4	0.0064					
5% F	2.1	0.00062	5.05	0.003	0.002	0.002	0.0006
5% M	2.2	0.0047					
5% C	2.3	0.0069					

* F = Fine; M = Medium; C = Coarse.

* Chemical analyses by courtesy of the Aluminum Company of America, Research Laboratories.

will not be reproduced here⁽²⁹⁾. All of the alloys were annealed for a total of 96 hours at 578°K prior to testing.

The plastic properties of the dispersion alloys for a series of test temperatures are recorded in Figs. 10A to 10D. For purposes of comparison, analogous data for an alpha solid solution having the same mean grain diameter and copper content of 0.19 atomic % as the matrix in the dispersion alloys are given in Fig. 11. The data recorded in Fig. 10 are observed to be in general agreement with the postulate by Gensamer that the plastic properties of dispersed alloys are dependent primarily on the mean free path between the hard particles embedded in a soft ductile matrix.

Cross comparison of Figs. 10A to 10D reveal that the finer dispersions exhibit higher values of initial deformation strength and also slightly higher rates of strain hardening for all temperatures below 380°K. Moreover, comparison of these data with Fig. 11 reveal that the initial deformation strength and rate of strain hardening is appreciably greater for the dispersion alloys than for the supersaturated alpha solid solutions.

Since the stress-strain curves for the alloys under investigation here appear to be homologous, it might be anticipated that the deformation strength at some specified strain should depend on the volumetric mean free path between the CuAl_2 particles. Although the qualitative trend of dependence of the deformation strength on the mean free path was observed, the σ vs. logarithm V.M.F.P. relationship was curvilinear. When, however, the data were replotted as the logarithm of σ vs. logarithm of V.M.F.P., a linear relationship was obtained with a narrow scatter band as shown in Fig. 12. These correlations reveal that below 380°K, the stress-strain curves are homologous. In a single homologous series of alloys, the deformation strength will be a function of any suitable measure

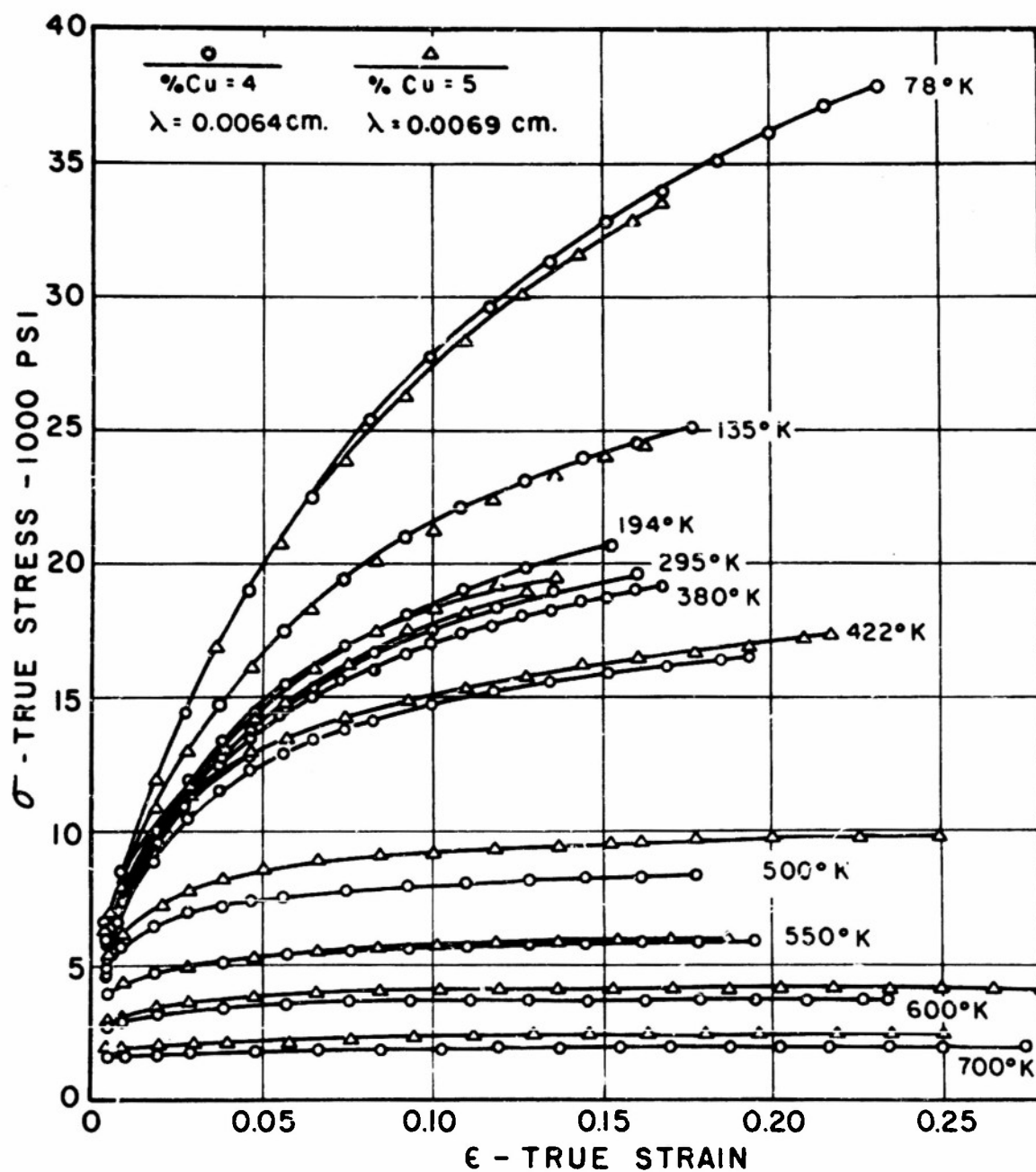


FIG.10A STRESS STRAIN CURVES FOR 5% COARSE AND 4 % COARSE Cu-AI DISPERSION ALLOYS

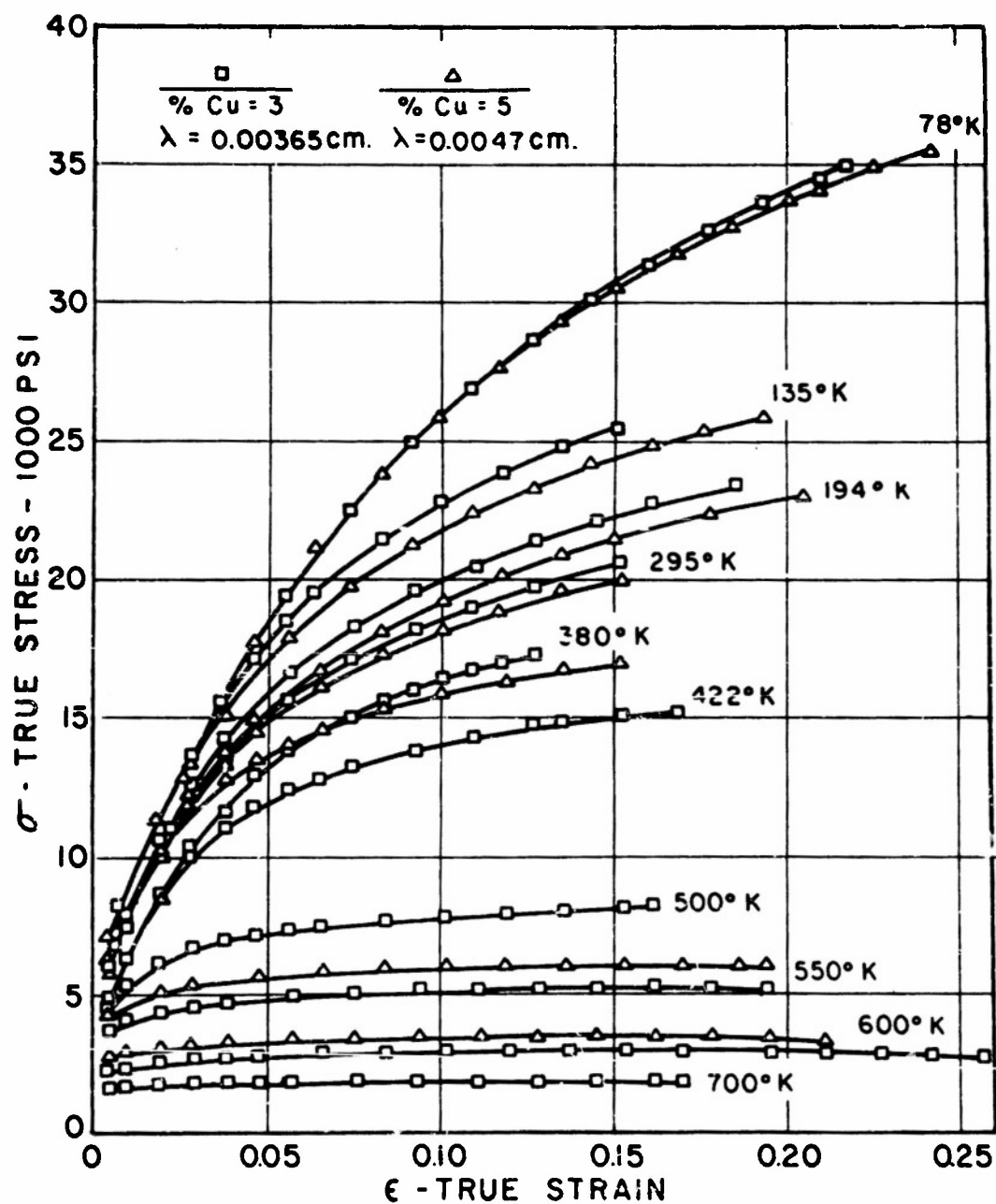


FIG. 10B STRESS STRAIN CURVES FOR 3% COARSE AND 5% MEDIUM Cu-Al DISPERSION ALLOYS

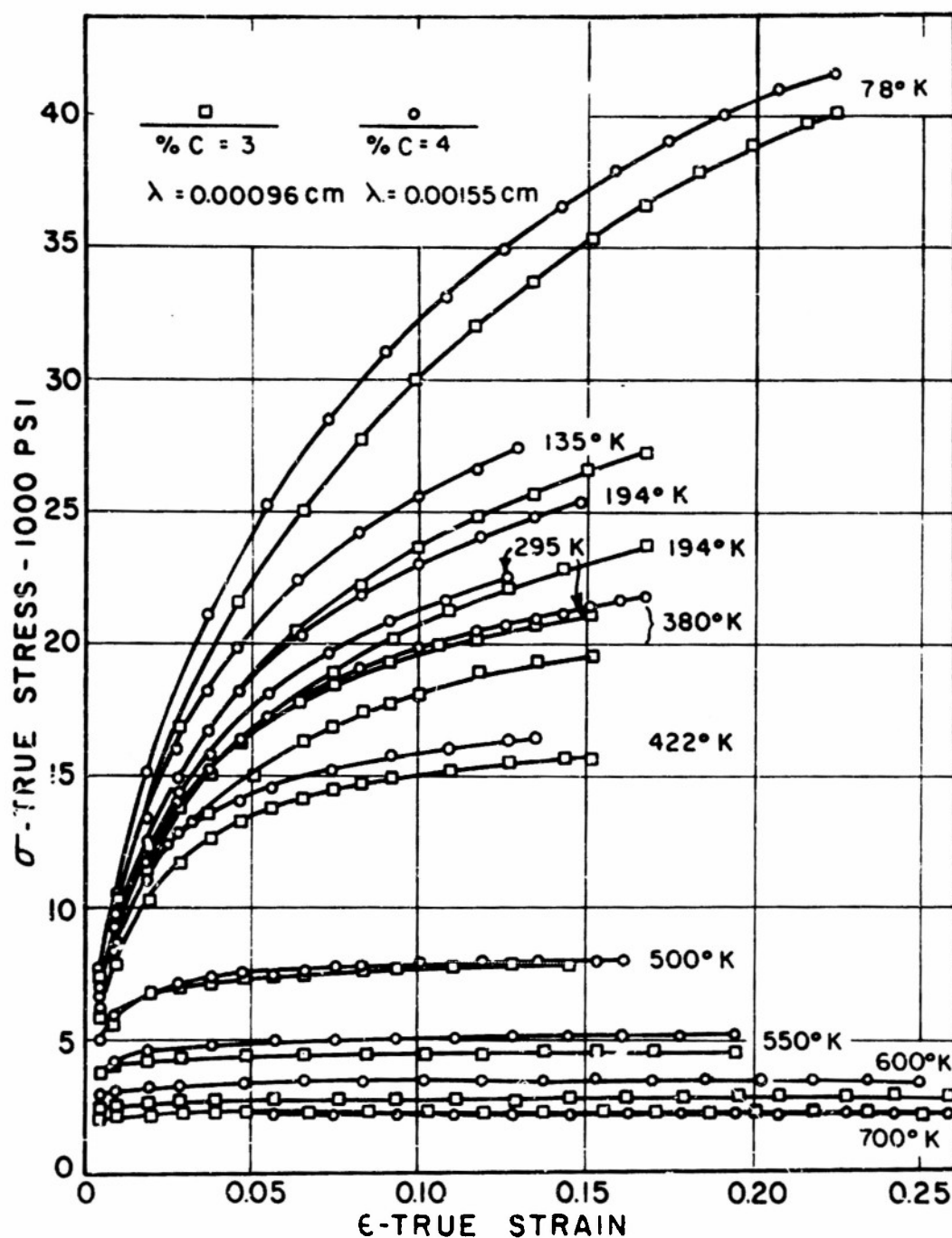


FIG. 10C STRESS STRAIN CURVES FOR 4% MEDIUM AND 3% MEDIUM Cu-AL DISPERSION ALLOYS

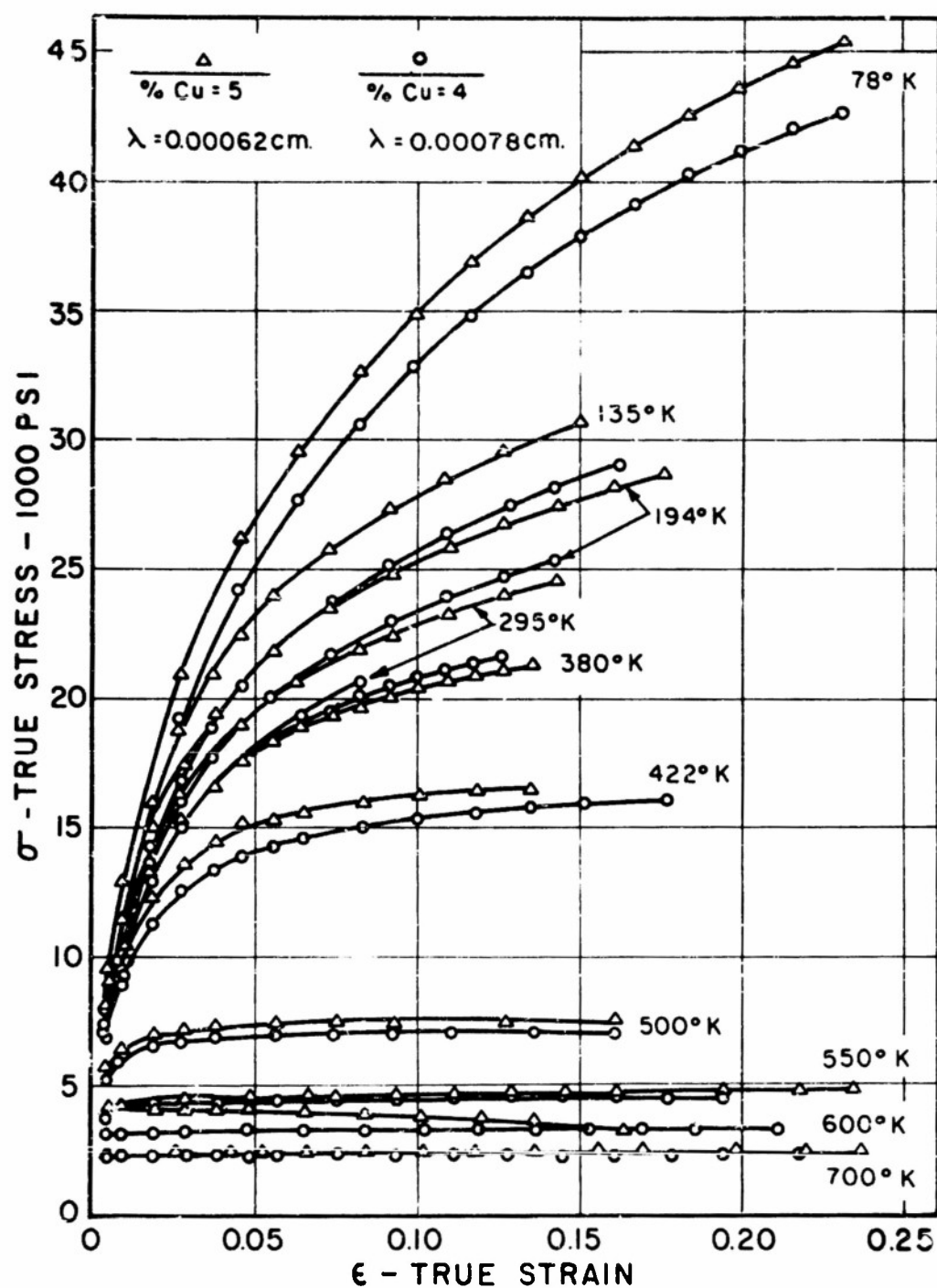


FIG. 10D STRESS STRAIN CURVES FOR 5% FINE AND 4% FINE Cu-Al DISPERSION ALLOYS

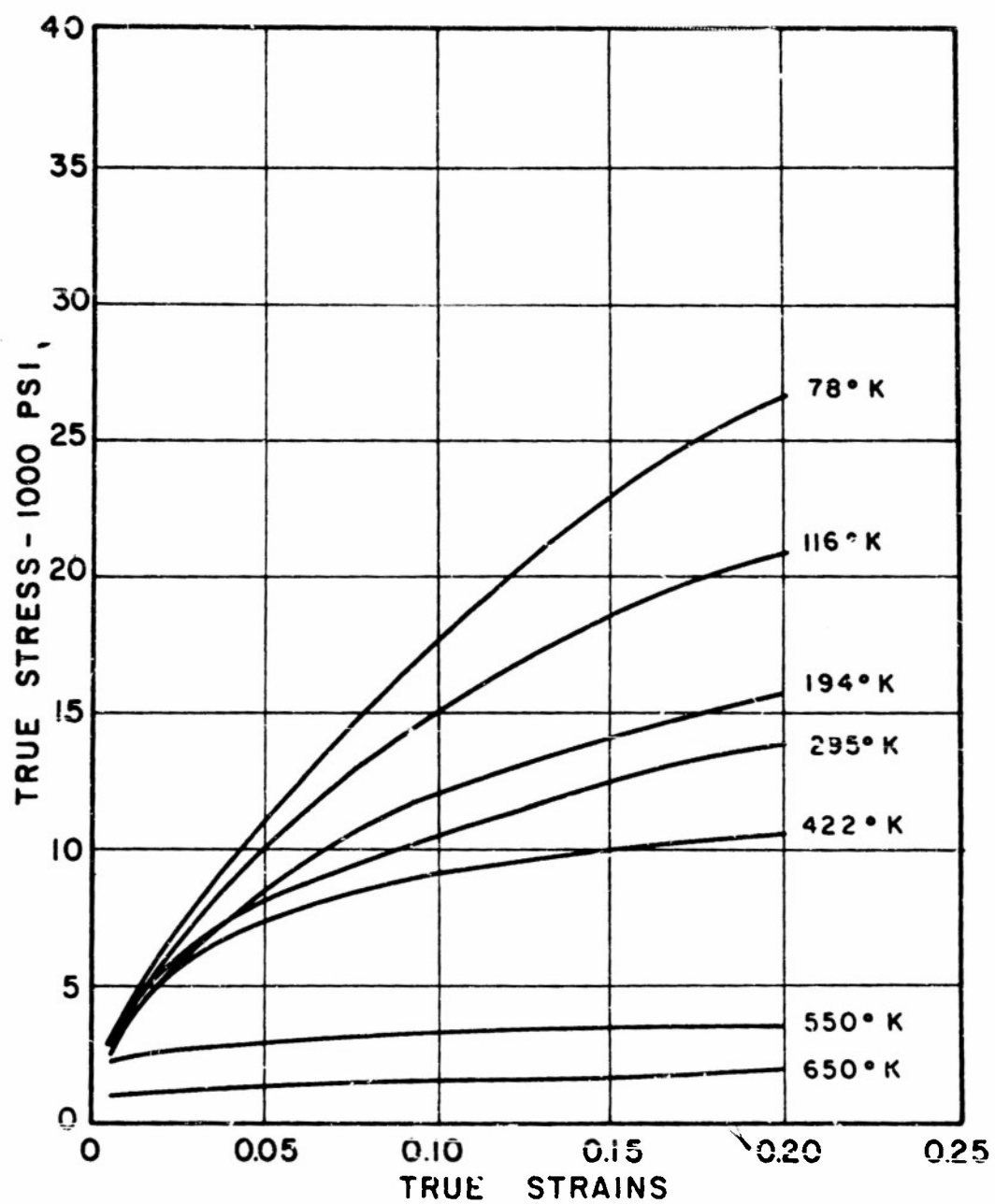


FIG.II STRESS STRAIN CURVES FOR Al-Cu SOLID SOLUTIONS CONTAINING 0.194 ATOMIC % COPPER

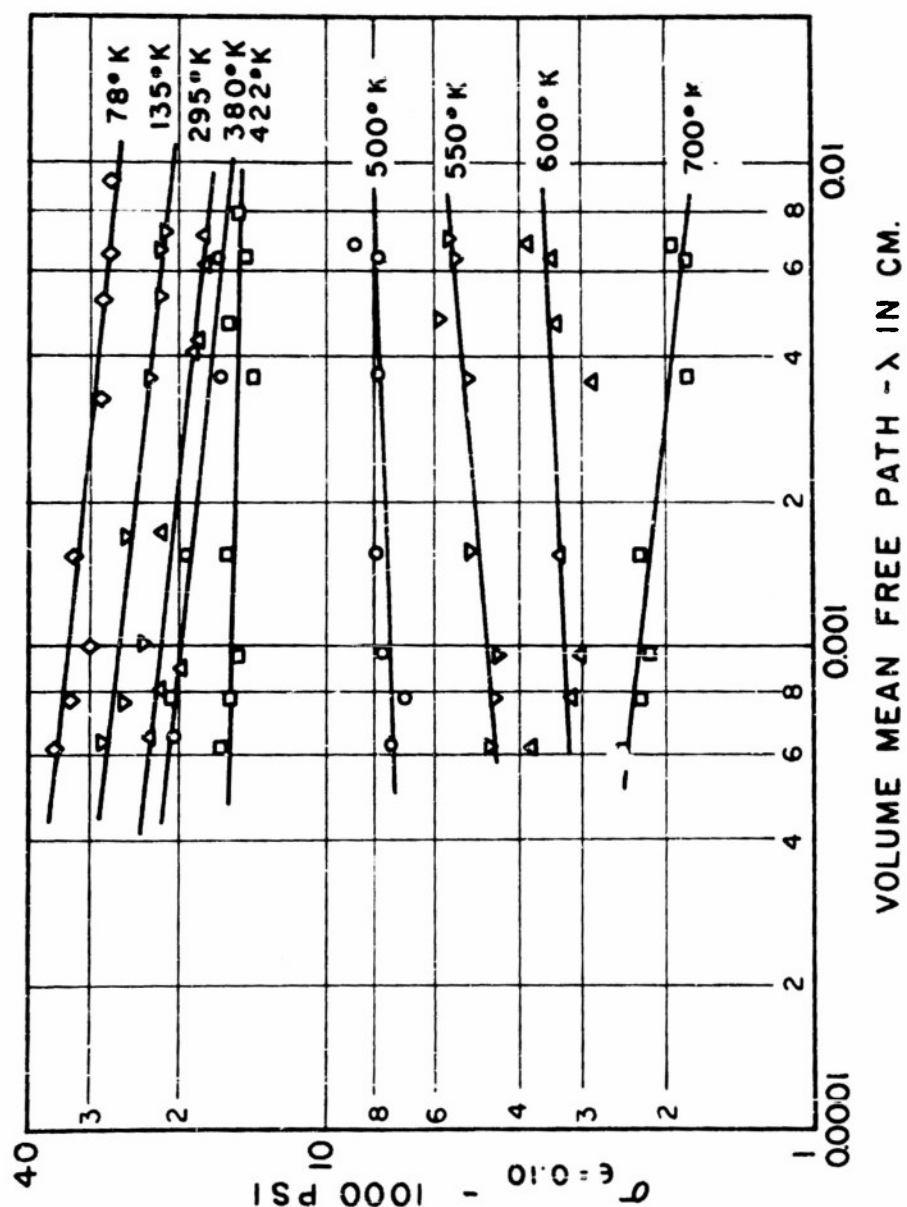


FIG.12 EFFECT OF VOLUME FREE PATH ON THE DEFORMATION STRESS
OF Al-Cu ALLOYS AT A PLASTIC STRAIN OF $\epsilon = 0.10$

of the dispersion. Consequently from the pure phenomenological viewpoint, no special significance can be ascribed to the fact that the deformation strength correlates with the mean free path between dispersed hard particles in a given homologous series of alloys.

Other empirical relations such as that by Unckel⁽³²⁾ have been used for representing the plastic properties of alloys containing dispersions:

$$\sigma = A e^{-B\lambda} + C \quad (5)$$

where σ = deformation stress
 B = constant
 λ = mean free path
 $A+C$ = strength of hard particles
 C = strength of matrix.

Although this equation does have the merit of predicting finite strengths for both the hard and ductile constituents, extrapolations to the strength of the ductile phase at low temperatures are approximately 50% in error.

In the temperature range from 422° to 600°K the slope of the stress-mean free path curve is the opposite of that encountered at the lower testing temperatures.

These data might at first appear to be somewhat unexpected and anomalous. The reason for this observation, however, is clarified by the typical results for the coarse and fine dispersions of the 5% Cu alloy shown in Fig. 13. Evidently the alloy having the more finely dispersed structure begins to recover at a low temperature. And somewhere above 422°K it recovers more rapidly than the alloy having the more coarsely dispersed structure. But at 700°K the recovery rate for both dispersions is so great that they are always in the almost completely recovered state.

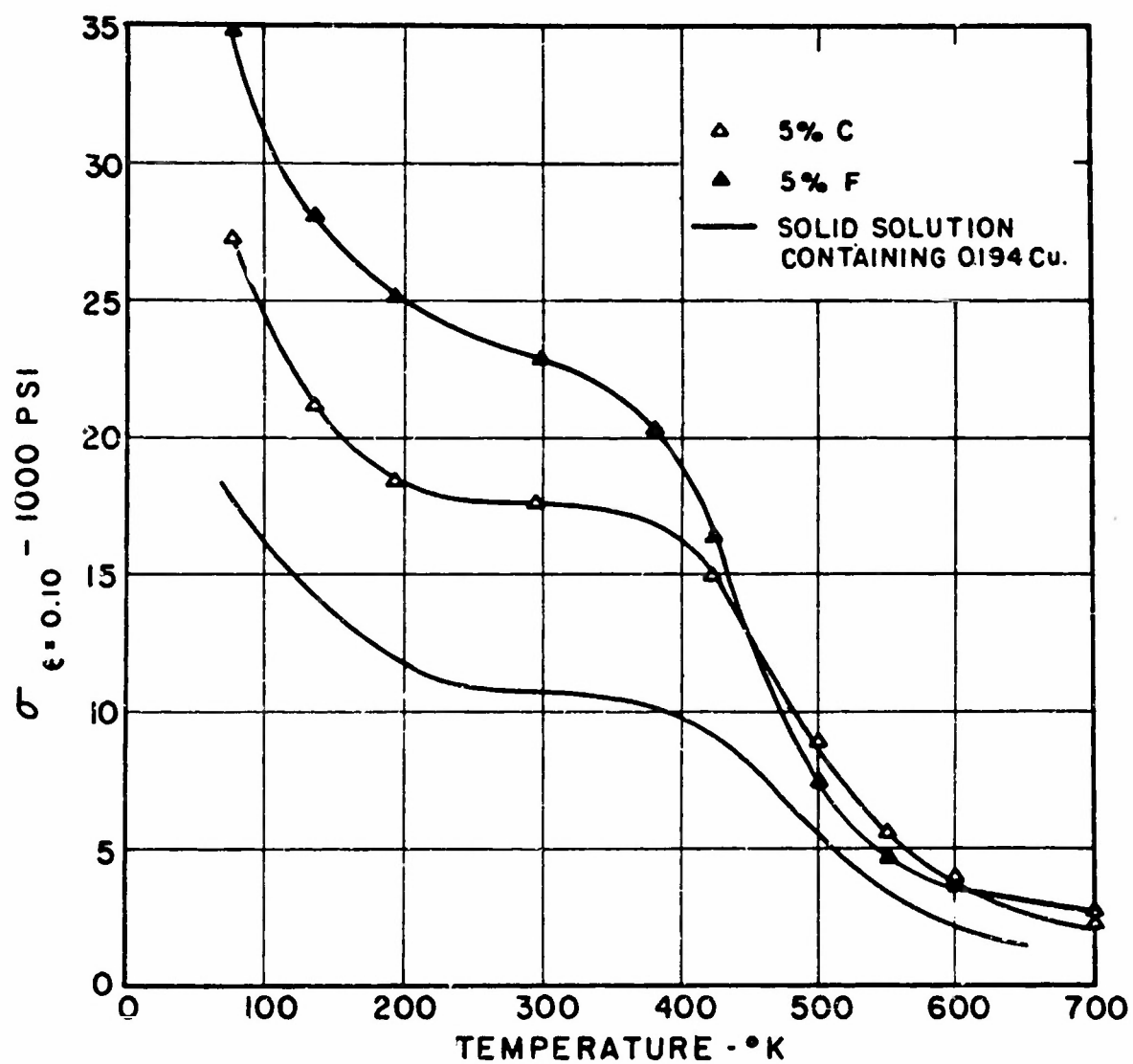


FIG.13 EFFECT OF TEMPERATURE ON THE DEFORMATION STRENGTH OF A 5% Cu-Al ALLOY AND A SOLID SOLUTION Al-Cu ALLOY AT A STRAIN OF $\epsilon = 0.10$

Consequently from about 450°K to about 600°K the coarser dispersion exhibits a higher strength but at 700°K the normal trend of increase in strength again prevails with refinement of dispersion.

In view of the objectives of Dr. Hart's⁽³³⁾ analysis of theories of dispersion hardening no attempt will be made here to review this phase of the subject.

SUMMARY

The deformation processes occurring in alloys composed of several ductile phases might be viewed as a simple extension of the processes that occur in single phase polycrystalline aggregates. Since no completely satisfactory theory has yet been established for the single phase polycrystalline analogue, no satisfying method of predicting the plastic properties of alloys containing two or more ductile phases has yet been formulated. The problems of major importance in extending our knowledge in these fields center about grain size effects, orientation effects, simultaneous satisfaction of stress equilibrium and continuity over the grain boundary regions, and heterogeneous deformation and stressing over a single grain.

The plastic behavior of alloys containing hard intermetallic phases dispersed in a soft ductile matrix has been shown to be primarily dependent on the mean distance between the hard particles. Excluding those cases where the brittle phase is continuous, dispersion hardening is only slightly sensitive to the type of dispersion be it lamellar or spheroidal. Although our present knowledge on dispersion hardening is yet largely empirical, some progress in applying the dislocation theory to this problem has been achieved by Dr. Hart.

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